

The Effect of Vanadium on the Phase Transformation of the 82B Steel

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Abstract

The effect of vanadium on the phase transformation and performance of 82B steel was studied. Continuous cooling transformation experiments and tensile tests were carried out by the dilatometer and tensile testing machine. The experimental results showed tensile strength and hardness enhanced as the vanadium content and cooling rate increased, and the formation of carbide network was encouraged by vanadium. Centre segregation was found in 82B steel with vanadium through component analysis. It was shown that the presence of vanadium in 82B steel refined the pearlitic interlamellar spacing, upgraded the grade of network carbide, promoted the generation of martensite and caused segregation and so on, and the corresponding mechanisms were analyzed. The research could improve the production process and optimize composition of 82B wire rod, providing guidance for the production practice.

Keywords

Micro-alloying Element; Vanadium; 82B Wire Rod; Phase Transformation; Performance

Introduction

With the development of the building materials industry, ordinary carbon steel wire rod can no longer meet the needs of users. Therefore, the high strength wire rod has been widely used such as 82B wire rod that is mainly used in the field of high-strength construction, usually drawn or twisted to be steel wire and strand. The rod should have sufficient high strength, stiffness and better drawing performance. The major factors that influence these performances are the ratio of sorbite, the degree of carbide network and martensite content [1].

The performance of 82B wire rod was closely associated with its composition. It was observed that strong carbide forming elements had significant influence on pearlitic transformation. The strength of the steel with vanadium was much higher than the value calculated by the formula of Hall-Petch, which simply considered the contribution of grain refinement.

Scholars generally proposed that this increase of strength was caused by precipitation strengthening. A decrease in the transformation temperature, which is due to an increase in cooling rate, affects the level of precipitation strengthening. These precipitates are closely spaced and obstruct mobile dislocations, resulting in higher strength [2,3].

The main form of micro-alloying elements in steel is dissolved in the iron matrix or micro-alloying carbonitrides, and micro-alloying elements of various forms played different roles. Precipitation strength engendered by micro-alloy element vanadium enhanced the strength and hardness of material. And the segregation of vanadium could refine pearlitic interlamellar spacing [4]. However, 82B steel as hypereutectoid steel is easy to generate carbide network and martensite, which can severely reduce the ductility and toughness of steel. I.D.Mclvor [5] had done some research on the microalloying of wire rod. Strength was increased by precipitation strength and pearlitic interlamellar spacing was also refined through his research of adding vanadium to the steel with 0.8 wt.% carbon. But he didn't propose the effect of vanadium on the carbide network, and the vanadium caused a greater risk of generating martensite, which decreased the ductility. Thus it was unknown about the effect of adding vanadium to 82B steel on the performance in the actual production. This paper focused on studying the effect on the microstructure and performance of adding vanadium in 82B steel, and was purposed to provide the guidance for the practical processing.

The following proposed experiments were to provide a scientific basis for the Stelmor Controlled Cooling System (SCCS) for the high-speed wire rod production line [6]. In order to improve the product quality of the wire rod, this system has been installed and operated in good condition. A series of tests was conducted to derive more precise forecasting results and make the online model prediction results be agreement with

actual production.

Experimental Material

All samples in this experiment were collected from the production site. The serial numbers of the specimens were 1# and 2#. Their chemical composition was given in table 1.

TABLE 1 THE COMPOSITION OF 82B STEEL

element	1#(wt%)	2#(wt%)
C	0.82	0.82
Mn	0.75	0.75
P	0.014	0.014
S	0.001	0.001
Cr	0.22	0.22
Si	0.22	0.22
Cu	0.05	0.05
Ni	0.02	0.02
V	0	0.03

Experimental Procedure

CCT curves of the two specimens were employed to analyze the effect on the cooling characteristic and network carbide of 82B steel with micro-alloying element vanadium. Firstly, the 82B wires rods were machined to specimens with 4 mm diameter and a length of 10 mm which were heated at the rate of 10°C/s until it reached 900°C, then insulated for 10 min. After that, cooling respectively at the rate of 0.5, 1, 3, 5, 6, 7, 8, 9, 10, 11, 13, 15, 20°C/s.

The two specimens were kept under the same test conditions as much as possible, then polished and etched in the 4% nital, and metallographic structure could be observed by OM. The continuous cooling curves (CCT) could be drawn by combination with metallographic observation, hardness test and measuring the interlamellar spacing [7]. Then a systematic analysis would be carried out to compare the cooling characteristic and carbide network of the two specimens. Additionally, complexity analysis of the original sample was done to analyze the effect on the segregation of 82B steel by adding micro-alloying element vanadium. The original sample was 12.5 mm diameter wire rod and taken respectively from the specimens of 1# and 2# at the area within 2 mm in the center and 1 mm near the outer surface of the wire rod.

Experimental Results and Analysis

Component Analysis

It was available and accurate by using chemical analysis about this analysis report. Instrumental

inaccuracy was $\pm 5\%$ and artificial calibration was employed, which made the inaccuracy smaller. The analyzed result was shown as table 2.

TABLE 2 THE REPORT OF COMPONENT ANALYSIS

No. element	1#surface (wt%)	1#core (wt%)	2#surface (wt%)	2#core (wt%)
C	0.81	0.82	0.80	0.84
Mn	0.78	0.74	0.70	0.80
P	0.012	0.013	0.013	0.016
S	0.001	0.001	0.001	0.0014
Cr	0.23	0.20	0.16	0.25
Si	0.23	0.25	0.28	0.22
Cu	0.048	0.051	0.042	0.056
Ni	0.022	0.020	0.020	0.026
V	0	0	0.026	0.034

From the analysis report, the segregation in 2# specimen was more significant than 1#. There was basically no segregation in 1# specimen and the composition of the surface was almost the same as the center. In 2# specimen, segregation of carbon, manganese, chrome and nickel toward the core was obvious for the presence of vanadium. This suggested that the synergy of vanadium with other micro-alloying elements was very strong. Segregation was locally generated in the core of the wire, and the local composition of the steel was aberrated because of segregation. In the center somewhere the carbon content was markedly higher than eutectoid steel and unexpected thick cementite network was generated during continuous cooling.

In the micro-alloying steel, micro-alloying elements always cooperated with each other, forming carbide, nitrides or carbonitrides with carbon and nitrogen. Highly dispersed carbonitride particles was generated by micro-alloying elements, which enabled effective pinning of the austenitic grain boundary, prevented the migration of the austenitic grain boundary, and inhibited the growth of austenitic grain. These compounds dissolved under high temperature and precipitated under low temperature. The growth of the initial austenitic grain was inhibited by the precipitation during the reheat progress. When heated, the solubility of vanadium improved with the increasing temperature and manganese content. Meanwhile, recrystallization and following grain growth were prevented by the precipitation when rolled. When rolled under high temperature, the precipitation of the carbonitride was not evident and the rate of precipitating was not fast, because of disappearing of the dislocation and distortion generated by dynamic recrystallization and deformation. On the contrary, when rolled at low temperature, the

recrystallization did not happen in austenite and the imperfection increased, which was encouraged to form the network carbide. In the cooling process, the surface temperature of the wire was lower than the center and the segregation mainly related to the difference of temperature.

CCT Curves of the Two Specimens

By combining thermal expansion method with metallographic hardness method to draw the austenite continuous transition curve, the expansion curves could firstly be drawn, and then the critical temperatures of each transformation at different cooling rate were determined. Based on requirement of the national standard YB/T5128-93, CCT curves were drawn and given in Fig. 1.

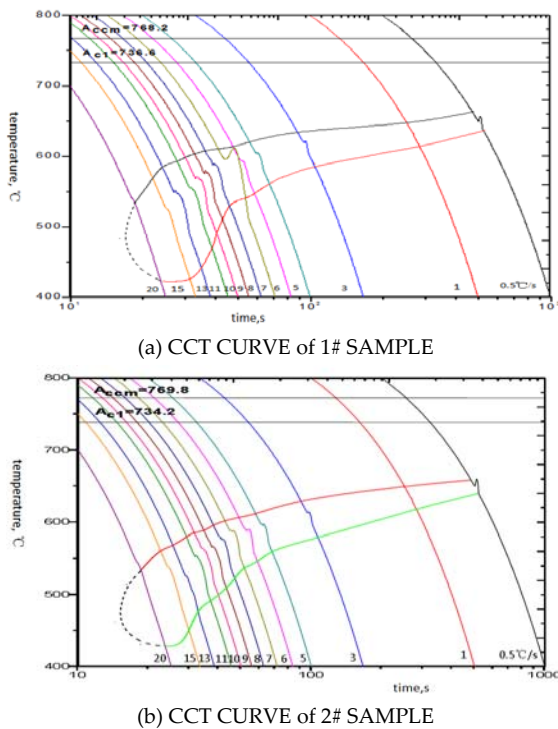


FIG. 1 CCT CURVE OF 82B STEEL

Combined CCT curves with metallographic pictures, specific analysis could be carried out [8,9]. Austenite was stable before the sample was cooled to the temperature of A_{c1} . As the temperature dropped below the A_{c1} , undercooling degree was different as the varying cooling rate (CR), and different microstructures were achieved. Pearlitic transformation would only occur when the cooling rate was very low. With the increase of cooling rate, the pearlitic transformation was not complete and martensitic transformation occurred. With further increase of CR, only martensitic transformation occurred. Because the incubation period of bainite was much longer, the bainitic transformation did not occur.

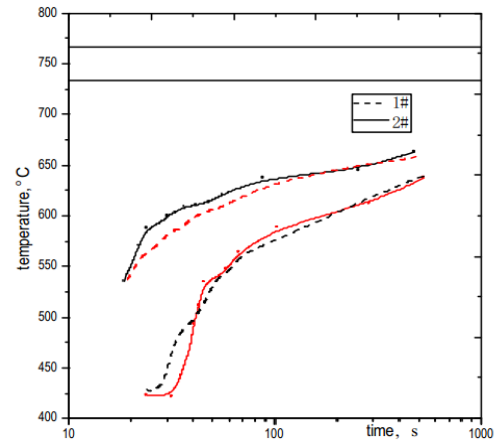


FIG. 2 COMPARISON OF THE TWO CCT CURVES, DASH LINE WAS 1# AND SOLID LINE WAS 2#.

In 1#, when the CR increased from 0.5 to 1°C/s, lamellar pearlite was clearly observed. With the gradually ascending of CR, the interlamellar spacing of pearlite decreased gradually. The pearlite lamellar morphology was shown in the Fig. 3. Martensite was found at the cooling rate of 5°C/s by the observation of microstructure and the test of hardness. When the CR was 13°C/s, the lath-shaped martensite was obvious. But the increment was not evident when the CR was under 13°C/s. The martensitic production significantly increased as the CR was above 13°C/s. Because the CR was greatly faster than the critical generating cooling rate of martensite so that it would be helpful to the formation of martensite.

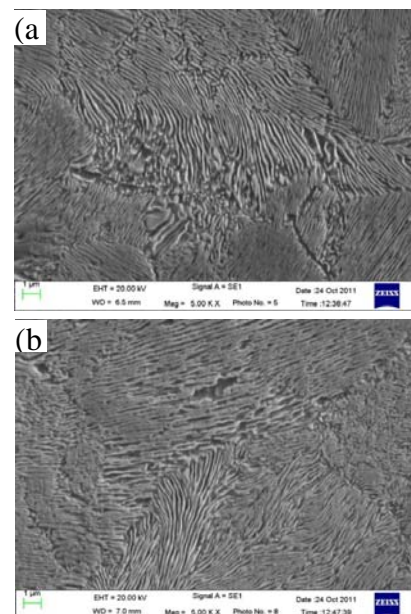


FIG.3 SEM METALLOGRAPHIC PICTURES OF THE SAMPLES
(a) 1#-1°C/s, (b) 2#-1°C/s

In 2# specimen, the martensite was obvious when the CR reached 10°C/s, and the martensitic production significantly increased as the CR above 10°C/s, which indicated that the fraction of martensite in 2# was

larger than that in 1# at high cooling rate, that is to say, vanadium facilitated the generation of martensite in 82B steel at high cooling rate as 10°C/s . The microstructure of 1# and 2# at 10°C/s was shown in Fig.4.

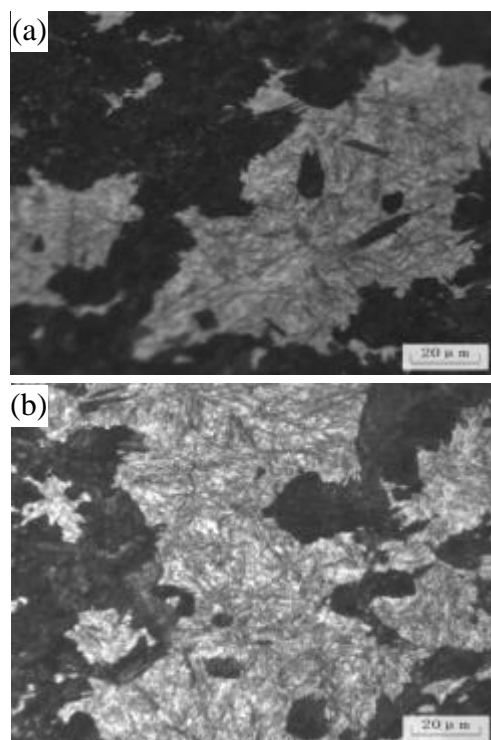


FIG. 4 THE MICROSTRUCTURE OF THE SAMPLES AT 10°C/s ,
(a) 1#- 10°C/s , (b) 2#- 10°C/s

The diffusion coefficient of vanadium was far less than that of carbon in austenite, and the whole phase change process was deferred by vanadium which delayed the nucleation and growth of carbide during the pearlitic transformation. Vanadium was strong carbide forming element and it easily caused center segregation combining other elements such as chromium, manganese. The pearlitic transformation in the center of the 82B wire rod was significantly delayed by the segregation, which caused pearlite to be generated in the surface of the wire rod while brittle martensite was formed in the center.

Assessment of Network Carbide's Grade

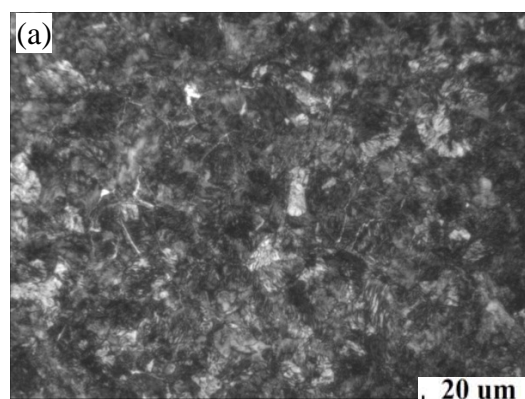
Carbide is generally very hard and also brittle. When it was transformed into the network carbides, the steel quality would become brittle and the toughness would become worse. Pencil-tip shaped fracture of 82B wire rod was the damage typically caused by the center of retention carbide. Pencil-tip shaped fracture was a shear fracture which occurred from the axis parts. The segregation of Cr and Mn led to the generation of a small amount of martensite in the axial core. Carbon

segregation caused the appearance of carbide network in the axial core. In the process of drawing to be strands, the part of the wire rod where carbide network existed was difficult to bear the processing load, then micropores formed, finally these micropores grew up together. At last this part of the wire rod would fracture under shear stress after the micropores' transition to the normal region. Therefore carbide network must be prevented in the range of standard. The standard of carbide network grade was expressed in GB/T 24242.1-2009, and the result of the two samples was shown in the table 3.

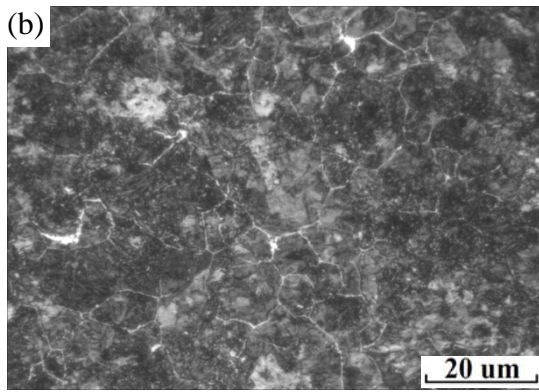
TABLE 3 COMPARISON OF SAMPLES' CARBIDE NETWORK GRADE

CR/ $^{\circ}\text{C}$	1#	2#
0.5	1	1.5
1	1	2
3	1.5	2.5
5	2	3
6	2	2.5
7	2	3
8	3	4
9	2.5	2.5
10	2	3
11	1.5	1.5
13	2	2
15	1.5	1.5
20	-	-

From the above table, the relationship between carbide network grade and cooling rates could be derived and compared [10]. The two specimens' grades of carbide network were similar when the CR was from 0.5°C/s to 5°C/s . While the CR was from 6°C/s to 10°C/s , and the grade of carbide network in 2# was 0.5 to 1 grade higher than 1#. The grade of carbide network decreased when the cooling rate was 11°C/s for the reason that a large number of martensite was generated and the quantity of pearlitic was reduced. Therefore the carbide network was mainly distributed along the pearlitic grain boundaries.



(a) 1# MICROSTRUCTURE AT 8°C/s



(b) 2# MICROSTRUCTURE AT 8°C/s

FIG. 5 METALLOGRAPHYS OF THE SPECIMENS(1# AND 2#)

The OM pictures of microstructure at 8°C/s were shown in Fig.5. Carbide network was clearly observed in 2# microstructure. And the pearlitic interlamellar organization distributed widely. It can be seen that these samples could identify the effect of vanadium on the network carbide.

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From the above comparison, the generation of carbide network was enhanced by adding vanadium. The grade of carbide network has been raised by 1 grade in the decomposition of austenite to pearlite as the effect of vanadium, which was different from the result of A.M. Elwazri [11] and co-workers who raised that carbide network was weakened by adding vanadium as carbide network to be blocked by vanadium carbide. For the solubility decreasing of carbon in the austenite during cooling, the atom in the phase boundary or grain boundary had larger free energy and movement space so that precipitation of the carbide happened along the boundary. Therefore, it was formed as the shape of net. And carbide network was mainly generated in the center of the wire rod, which indicated that it is related to the cooling rate and segregation. When the wire rod of 12.5 mm diameter was produced, there was a great temperature gradient from surface to center. That is to say, segregation and abnormal structure were generated for the slow cooling rate in the center. While vanadium was strong carbide forming element and it also contributed to the generation of martensite. So that the carbide network in 2# specimen was more than that in 1#, and the content of martensite in 2# was also more than that in 1#.

Critical Temperature and Microstructure

The critical temperature of each transformation was derived through analyzing the expansion curves. Also the microstructural types were determined through metallography combined with microhardness, as shown in Table 4.

With the increase of CR, the starting temperature of pearlitic transformation dropped gradually. This change was mainly due to the increase of CR, the undercooling degree also increased, and the ability of carbon diffusion decreased, thus delaying the occurrence of phase transitions, causing a reduction of the phase transition temperature. As the decreasing phase transition temperature, it contributed to the generation of a large number of martensite. Simultaneously, the time of pearlitic transformation was gradually reduced with the ascending of cooling rate. When the rate of transformation increased to a certain extent, residual austenite was generated due to the space constraints of the residual part of the austenite. Part of the starting temperature of martensitic transformation was not measured because too little martensite was generated.

Although martensite generation enhanced the hardness, toughness was greatly reduced at the same time. As the 82B wire rod was used to produce high-strength prestressed steel wire and strand, which required excellent toughness, the generation of martensite must be strictly controlled.

Generally, the transition temperature ranging from austenite to sorbite was from 650°C to 550°C. In this range, the lower the transformation temperature was, the finer the interlamellar spacing was derived, and the faster the cooling rate was, the higher the strength was derived. On the whole, the transformation starting temperature of 2# was lower than that of 1# and the transformation ending temperature of 2# was higher than that of 1#, namely having larger degree of undercooling and shorter range of transformation temperature. The transformation starting temperature was sharply reduced for the steel with vanadium. Moreover, the nucleation of cementite was effectively promoted by vanadium and transformation time was shorter, which was better to refine the interlamellar spacing of pearlite. That was why vanadium was often used as industrial surfactant.

Microstructural Changes in Volume and Performance

The volume fraction of microstructure after transformation was measured through the standard method, as shown in table 5.

TABLE 4 CRITICAL TEMPERATURE AND MICROSTRUCTURE

sample	CR(°C/S)	A _{cl} (°C)	A _{cm} (°C)	P _s (°C)	P _f (°C)	M _s (°C)	A-P transformation time(s)	microstructure
1#	0.5	736.6	768.2	663.2	636.9	— —	52	P+NetC(1)
	1			645.1	612.3	— —	32	P+NetC(1)
	3			637	588.6	— —	16	S+NetC(1.5)
	5			621.1	563.6	— —	11	S+M+NetC(2)
	6			613.3	547.5	— —	11	S+M+NetC(2)
	7			610.5	540.4	— —	12	T+M+NetC(2)
	8			608.6	534.7	— —	9	T+M+NetC(2)
	9			603.8	511.6	— —	10	T+M+NetC(2.5)
	10			599.6	493.8	194.9	10	T+M+NetC(2)
	11			592.6	470	195	11	T+M+NetC(1.5)
	13			587.8	442.8	199.1	11	T+M+ A _R +NetC(2)
	15			570.4	421.7	198.7	10	T+M+ A _R +NetC(1.5)
	20			535.1	422.5	197.3	5	T+M+ A _R
2#	0.5	734.2	769.8	657.8	639.1	— —	38	P+NetC(1.5)
	1			649.5	617.8	— —	31	P+NetC(1)
	3			631.2	577.6	— —	18	S+NetC(1.5)
	5			610.2	562.2	— —	9	S+M+NetC(2)
	6			605.9	543.5	— —	10	S+M+NetC(2.5)
	7			601.4	538.2	— —	9	T+M+NetC(3)
	8			595.5	521.2	— —	9	T+M+NetC(3)
	9			587.1	507.8	— —	9	T+M+NetC(2)
	10			585.6	498.3	183.1	9	T+M+NetC(3)
	11			578.9	490	183.1	8	T+M+NetC(1.5)
	13			566.8	480	193.9	7	T+M+ A _R +NetC(2)
	15			561.5	427.5	194.5	6	T+M+ A _R +NetC(1.5)
	20			532.9	428.3	192.5	5	T+M+ A _R

Note: P_s-the starting temperature of pearlitic transformation; P_f- The end temperature of pearlitic transformation; M_s-the starting temperature of martensitic transformation; P-lamellar pearlite; S-sorbite; T-troostite; M-martensite; A_R-retained austenite; NetC-carbide network, and figures in brackets is the level.

TABLE 5 THE VOLUME FRACTION AND HARDNESS OF MICROSTRUCTURES

CR(°C/s)	magnification	pearlitic average vol%		martensitic average vol%		white area's hardness HV		black area's hardness HV	
		1#	2#	1#	2#	1#	2#	1#	2#
0.5	200	100	100	0	0	— —	416.5	449.9	467.6
1		100	100	0	0	— —	427.5	456.7	481.4
3		100	100	0	0	— —	461.0	461.2	493.6
5		100	100	0	0	562.9	— —	475.4	517.8
6		100	100	0	0	640.9	689.4	482.7	545.1
7		100	98.26	0	1.74	746.8	792.8	490.1	560.1
8		98.13	98.02	1.87	1.98	846.1	876.8	495.2	— —
9		96.76	98.76	3.24	1.14	920.8	950.8	500.4	— —
10		97.8	76.86	2.2	23.14	982.5	1080.6	— —	— —
11		95.74	50.79	4.26	49.21	1028.3	1174.3	— —	— —
13		90.71	48.34	9.29	51.66	1010.0	1231.8	— —	— —
15		34.95	44.88	65.05	55.12	1153.0	1288.5	— —	— —
20		10.57	7.56	89.43	92.44	1240.7	1386.8	— —	— —

When calculating the volume fraction, the inaccuracy was larger when CR was close to the critical cooling rate of the martensite generation. The reason is that there were only a few small lumps of martensite in the microstructure. And the microstructure was more uniform at the other cooling rates so that the inaccuracy was smaller. In addition, retained austenite distributing among the martensitic needle was scarce,

so its volume fraction could be calculated with martensite together [12].

According to the measured results, the hardness values increased with the cooling rate, and the white areas and the black areas had similar hardness. Combined with the metallographic picture of different cooling rate and SEM, they should be the same kind of microstructure, namely pearlite.

With the CR increasing from 0.5 to 5 °C/s in 1#, pearlitic layer was finer gradually, meaning that the degree of sorbitic was deeper. And the overall mechanical performance of the samples was also enhanced so that the stiffness increased gradually. Hardness was significantly improved until the CR reached 6 °C/s, and martensitic islands could be observed in the center. With the increase of the cooling rate, large pieces of sheet martensite and lath martensite were found and further increasing stiffness happened because of the enhancement of martensitic transformation. It can be seen that martensite generated temperature in 2# specimen was lower than that in 1#, which indicated that the generation of martensite was evidently impelled by adding vanadium. The hardness of 2# specimen was higher than that of 1#, for the effect of vanadium on the sorbite amount at low cooling rate, namely the effect of precipitation strength. At high cooling rate, it was the same situation for the effect of vanadium on the generation of martensite.

Table 6 shows the tensile strength and ductility of the sample. From the experimental results, tensile strength increased and ductility decreased by adding vanadium. And the same happened with increasing cooling rate.

TABLE 6 TENSILE STRENGTH AND DUCTILITY OF THE SAMPLE

Sample No.	Cooling Rate (°C/s)	Tensile Strength (MPa)	Ductility (%)
1#-1	0.5	1032	22
1#-2	1	1048	24
1#-3	3	1055	25
1#-4	5	1058	27
2#-1	0.5	1046	21
2#-2	1	1054	22
2#-3	3	1075	23
2#-4	5	1083	23

The strength of the steel ascended with the adding of the vanadium, which means the gradual increasing of the precipitation fraction. Additional strength caused by abundant vanadium was larger for the increasing carbon content. The generation of refining vanadium precipitation was promoted by this effect and more strength was provided. To avoid excess strength, it was necessary to control the composition subjected to the following restrictions [12] :

$$\%V \times \%C \leq 0.0026$$

Pearlitic Interlamellar Spacing

Another important examination index of 82B wire rod is degree of sorbite. The smaller the pearlitic interlamellar spacing was, the higher the intensity and the better plasticity were. When ferrite and cementite films are thin, there are more phase interfaces. Because

of the interfaces, the ability of resistance to plastic deformation grows under the external force, and ferrite and cementite films are very thin, which incites the increasing plastic deformation [13]. As 82B steel is expected to obtain the desired performance, strict requirements are needed on the degree of sorbite. For this purpose, the pearlitic interlamellar spacing needs to be calculated through the scanning photographs.

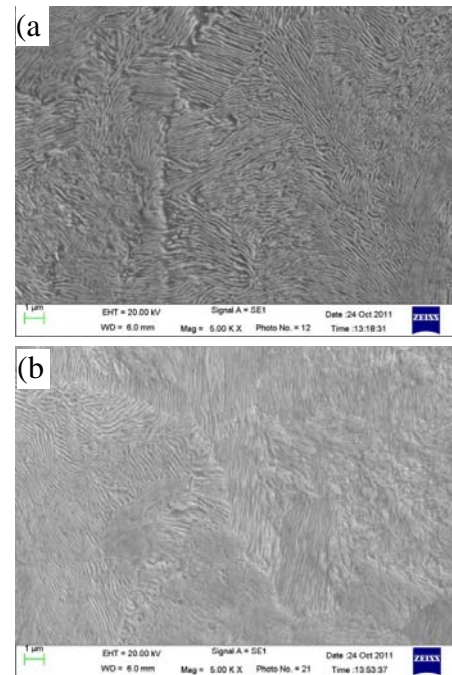


FIG. 6 SEM METALLOGRAPHIC PICTURES OF THE SAMPLES, (a) 1#-5°C/s, (b) 2#-5°C/s

Observed from the above photographs shown in Fig.3 and Fig.6, the lamellar spacing reduced gradually on the whole with the increasing CR. Lamellar microstructure was evenly distributed. The average lamellar spacing of pearlitic was derived by using SEM photography to measure the lamellar spacing. The mechanical properties were affected by the grain size. The material toughness increased with the decrease of the grain size. For high carbon steel, the pearlitic interlamellar spacing affected the mechanical properties of materials. It was found that the lamellar spacing under different cooling rates was not the same, and the interlamellar usually became smaller with the increase of the pearlitic cooling rate. The following table showed the calculated average pearlitic interlamellar spacing.

TABLE 7 THE AVERAGE PEARLITIC INTERLAMELLAR SPACING

CR(°C/s)	0.5	1	3	5	6	7
1#(μm)	0.26	0.22	0.17	0.14	0.11	0.09
2#(μm)	0.24	0.18	0.15	0.12	0.102	0.08

It could be found from the results that pearlitic interlamellar spacing decreased gradually with the

ascending cooling rate. The pearlitic interlamellar spacing in 2# specimen was thinner than that in 1# at each cooling rate. When the CR was 0.5°C/s, the degree of sorbitizing was very high and there was already a lot of troostite. With the decreasing interlamellar spacing, the mechanical properties of the 82B steel were improved.

In all the micro-alloying elements, vanadium was the element which was most suitable to generate stable and strong precipitation strength. Mainly because the dissolubility of VC was great, VC had strong ability to dissolve under high temperature. When the temperature was above 1000°C, VC existed in the steel as solid solution state. At the initial stage of pearlitic transformation, segregation was generated by the little vanadium in austenite, and the enrichment of carbon happened on the grain boundary. Then the nucleation driving force of carbide was enhanced. The trend of generating carbide was raised by the segregation of vanadium and carbon. Namely, the nucleation rate of cementite on the grain boundary ascended. And as the further nucleation and growth of cementite, the carbon around cementite was depleted. With the enlarging carbon-poor area, the growth of carbide was restrained. Then ferrite nucleated along the grain boundary around the precipitation of carbide, a single ferrite formed on the interface of cementite or austenite. When their geometric orientation was gradually built, ferrite and cementite would grow synergetically. This segregation was stronger as the increasing vanadium, which raised the nucleation of cementite, and then the pearlitic interlamellar was refined.

Zener [14] pointed out that the pearlitic interlamellar spacing corresponding to the fastest coarsening rate of pearlite was as the following formula shown, and interlamellar spacing under equilibrium conditions was $2S_c$.

$$2S_c = \frac{4\sigma V_m T_e}{\Delta H \cdot \Delta T}$$

Where σ was the specific value of interfacial energy between ferrite and cementite, V_m was specific value (supposing that V_m of each phase was equal), T_e was A1 temperature, ΔT was degree of supercooling, ΔH was latent heat of phase transition. It was not difficult to find from the above formula that the pearlitic interlamellar spacing was inversely proportional to the degree of supercooling. The higher the CR was, the greater the degree of supercooling was reached. Thus the pearlitic interlamellar spacing was reduced

with the increasing CR.

Meanwhile, vanadium was the element reducing the austenite phase region. A1 and A3 point would ascend with the addition of vanadium, which meant that steel with vanadium had larger degree of undercooling than the normal steel under the same isothermal temperature, namely having more finer interlamellar spacing.

Conclusions

The above experimental result and analysis were concluded as follows.

- (1) Complexity analysis was employed to investigate that the formation of carbide network was related to segregation. The segregation of carbon, manganese, chrome and nickel toward the core was obvious for the presence of vanadium. Because of segregation, the carbon content was markedly higher than eutectoid steel and unexpected thick cementite network was generated during continuous cooling.
- (2) The CCT curves of the two specimens were similar. The transformation initial temperature of 2# specimen was lower than that of 1# derived through the statistics of the critical point of the pearlitic phase transition, which indicated that the initial temperature was reduced and the degree of supercooling was enhanced by vanadium. Then the transformation time was shorter in the steel with vanadium.
- (3) The grade of the carbide network was rated. The carbide network in the 2# specimen was basically higher than that in 1# because vanadium added into 82B wire rod promoted the precipitation of carbide. As strong carbide forming element, vanadium segregated toward the center of the wire rod and contributed to the generation of carbide. Thus the grade of carbide network was raised.
- (4) The transformation rate of martensite in 2# specimen was faster than that in 1# specimen as the increase of the cooling rate. This suggested that the martensite transformation was accelerated by vanadium in 82B wire rod, and the assist was greatly especially at high cooling rate such as 10°C/s. Hardness and tensile strength increased and ductility decreased as the result of adding vanadium.
- (5) The interlamellar spacing of pearlite was calculated, and the average lamellar spacing of 2# was thinner than 1#. The lamellar was refined by the precipitation generated by vanadium. In addition, the nucleation of cementite was promoted by vanadium and the

symplastic growth of cementite and ferrite was accelerated, which could refine the interlamellar spacing.

(6) Known from the comparison of all the specimens derived from the test process, the 82B wire rod gained high combination property, while the cooling rate must be controlled under 8°C/s. Otherwise, the carbide network and martensite would be a tough trouble. Vanadium content should be not too high. For one hand, the cost of production increased. On the other hand, the carbon in steel would be carried off by the excess vanadium in the form of generating carbide, which reduced the carbon content in steel and then proeutectoid ferrite was formed, causing a reduced performance of the steel.

REFERENCES

- [1] Sun Ying, Yu Qing-bo, Analysis on microstructure and properties and defects of 82B steel wire rod[J]. Hot Working Technology, 2011,40(21):19-21;
- [2] M. Hajisafari, S. Nategh, H. Yoozbashizadeh, and A. Ekrami, Fatigue Properties of Heat-Treated 30MSV6 Vanadium Microalloyed Steel[J]. Journal of Material Engineering and Performance, 2013, 22(3): 830-839;
- [3] Ettore Anelli. Application of Mathematical Modelling to Hot Rolling and Controlled Cooling of Wire Rods and Bars. ISIJ International, Vol.32(1992), No.3, p.440-449.
- [4] Wang Yong-wei, Gui Mei-wen, Zhou Yong, et al. Effect of V and N micro-alloying on properties of high carbon steel 82B[J]. Transactions of materials and heat treatment, 2011, 1(32): P82-86.
- [5] I.D. McIvor, Microalloyed very low carbon steel rod [J]. Ironmaking and Steelmaking, 1989, 1(16): P55-62.
- [6] Yu_Wan-Hua, Chen Shao-Hui, Kuang Yong-Hai, Cao Kao-Chao, "Development of the 82B wire rod and Application of Online Stelmor Controlled Cooling SYstem" Applied Thermal Engineering, 2009, 29, 2949-2953.
- [7] Shuxiang Wang, Yu Wanhua, Chen Shaohui, Research to the controlled cooling technics of 82B wire rod[J]. Metal Product, 36(2010), 4, P.71-75.
- [8] Zhou Xu, Liancheng Zhao, The theory of metallic solid-state phase transition[M]. Science Press, Beijing, 2004, P.57-61.
- [9] Guang Xu, Wei Wang, Xinqiang Zhang, Measurement and drawing of the metallic material's CCT curve[M]. Chemical Industry Press, Beijing, 2009, P.119-120.
- [10] Yiqiang Sun, Guixian Gong, Lixin Wu, Comparison of detection methods in high-carbon eutectoid steel wire rod[J]. physical and chemical inspection, 44(2008), 10, P.544-546.
- [11] A.M. Elwazri, P. Wanjara and S. Yue. Effect of prior-austenite grain and transformation temperature on nodule size of microalloyed hypereutectoid steels [J]. Metallurgical and Materials Transactions, 36(2005): P2297-2305.
- [12] Feng Zhu, Chaoqun Wu, Weixue Tang, Research and application of applied materials to metallographic analysis system in metallographic quantitative analysis[J]. 4(2007), 1, P.318-320.
- [13] I.D. McIvor, Microalloyed high carbon steel rod[J]. Ironmaking and Steelmaking, 1989, 1(16): P49-54.
- [14] C. Zener. Kinetics of the decomposition of austenite[J]. Trans. AIME, 1946, 167: 550-596.